A critical analysis of existing models for plastic flow in Ni_3Al: Comparisons with transient deformation experiments

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A review of the intermediate temperature creep properties of Ni_3Al indicates that octahedral glide, the mechanism associated with the anomalous temperature dependence of yielding in this alloy, is exhausted during primary creep. Comparisons of this primary creep transient with the various models proposed to explain the anomalous yielding behavior suggest that additional transient experiments are needed to fully characterize octahedral glide in this alloy. Cottrell–Stokes type temperature change experiments, stress relaxation experiments, and deformation exhaustion/temperature drop tests have been conducted in an attempt to better characterize octahedral glide in Ni_3Al. The results of these transient experiments indicate that octahedral glide is only partially a thermally reversible process and that the formation of dislocation substructure, KW locks, plays an important role in determining the flow strength of this alloy. These experiments also suggest that flow in Ni_3Al should not be viewed as a viscous drag process, but can best be described as a "pure-metal-like" process involving the rapid motion of a small number of highly mobile dislocations. The stochastic nature of dislocation motion and the importance of substructure formation are emphasized and a description of octahedral glide that is consistent with the transient deformation experiments is proposed.

I. INTRODUCTION

The flow strength anomaly of Ni_3Al has been studied extensively and the major factors that contribute to the increase in yield strength with temperature have been identified. It is widely accepted that this anomalous behavior involves primary octahedral glide which is impeded by "cross-slip" onto the (010) plane, but the exact mechanism by which the formation of cross-slipped dislocation segments leads to the observed increase in flow stress has not been conclusively identified. Several dislocation models have been proposed to explain the yield stress anomaly.[1–10] However, for various reasons, none of them has been fully accepted as a complete description of octahedral glide in Ni_3Al.

In the work described in this paper, transient deformation experiments have been conducted to provide a basis for critically analyzing the various models of plastic flow that have been proposed. To date, the vast majority of experiments conducted on Ni_3Al have been constant strain-rate experiments. Recent observations of octahedral glide during the primary creep of Ni_3Al[11] have suggested the need to compare the results of these constant strain-rate experiments with observations of transient deformation in Ni_3Al. Deformation transients associated with changes in temperature, stress, and strain-rate have previously been used to determine the controlling mechanisms of deformation in various metals and alloys.[12–14] Here we use similar approaches to study the characteristics of plastic flow in Ni_3Al.

This paper includes a review of the models that have been proposed to explain the anomalous yielding behavior, an explanation of the transient experiments that have been conducted, a discussion of the results of these experiments including comparisons with the proposed models, and, finally, a general description of octahedral glide in Ni_3Al that is consistent with the present observations.

II. BACKGROUND

Dislocation motion in ordered Ll_2 structures occurs on {111} planes, just as it does in ordinary fcc metals, but the Burgers vector (**b**) is twice as large in the ordered compound. The elastic strain energy associated with this large **b** serves as a driving force for the dissociation of the unit dislocation into two or more partial

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dislocations. Pair potential calculations of Yamaguchi *et al.*[15] and weak beam transmission electron microscopy (TEM) observations by Sun and Hazzledine[3] and Baluc, Karnthaler, and Mills[16] indicate that the core of a unit dislocation for octahedral glide in Ni_3Al is actually a planar, ribbon-like structure containing four Shockley partial dislocations connected by an anti-phase boundary (APB) and two complex stacking faults (CSF's).

Glide is an easy process as long as the dislocation core remains planar, but the dislocation core in Ni_3Al is not always planar. The leading superpartial screw dislocation for primary octahedral glide, $\mathbf{b} = a/2[\bar{1}01]$, is free to cross-slip from the (111) plane onto the (010) plane. Atomistic studies by Yamaguchi *et al.*[15] and Paidar *et al.*[17] and high resolution electron microscopy (HREM) observations by Crimp and Hazzledine[18] and Mills[19] have shown that superpartial dislocations cannot expand easily on the (010) plane and will always dissociate into Shockley partial dislocations which lie on an octahedral plane. As a result, the motion of screw dislocations in the cross-slipped configuration is severely impeded. Continued motion on the (111) plane would require dragging of the APB, and glide on the (010) plane would require constriction of the Shockley partial dislocations.

A. Kear–Wilsdorf locks

Kear and Wilsdorf[20] were the first to observe long, relatively straight, screw dislocations in deformed specimens of ordered Cu_3Au. They suggested that these long screw dislocations had cross-slipped to the (010) plane and become locked. For this reason, dislocations in the cross-slipped configuration are commonly referred to as Kear–Wilsdorf locks (KW locks). Kear[21] used the formation of KW locks to account for the anomalous temperature dependence of work hardening in Cu_3Au; the work-hardening rate increases with temperature at intermediate temperatures. He reasoned that as a dislocation loop expands on the (111) plane there is a finite probability that the screw segments of the loop will cross-slip and become locked. Since the probability of this thermally activated cross-slip event increases with increasing temperature, it follows that the anomalous work-hardening rate can be explained in terms of the formation of KW locks. Based on the TEM observations of Kear and Hornbecker,[22] Copley and Kear[23] explained the work hardening of Ni_3Al single crystals in a similar manner. They showed that the formation of KW locks leads to strain hardening for single slip orientations and that the temperature dependence of the cross-slip event results in anomalous strain hardening in Ni_3Al as well as Cu_3Au. They did not, however, relate the formation of KW locks to the anomalous temperature dependence of the yield strength.

B. Takeuchi and Kuramoto's model

Takeuchi and Kuramoto[1] were the first to relate the formation of KW locks to the observed increase in yield strength with temperature through a reduction in the mobility of screw dislocations. Their model was based on observed violations of Schmid's law for the critical shear stress (CRSS) of Ni_3Ga, and it included a description of the frequency of cross-slip which is dependent on both temperature and the resolved shear stress on the (010) cube cross-slip plane. They proposed that when a screw dislocation glides on the (111) plane, small segments of it will cross-slip onto the (010) plane and form KW locks, but that the portions of the dislocation that remain on the (111) plane will bow out between the KW locks and eventually break away from them, Fig. 1. By assuming that for each temperature there exists an equilibrium number of locks determined by the frequency of cross-slip, Takeuchi and Kuramoto were able to derive an expression for the critical stress required to maintain dislocation motion:

$$\tau_c = \tau_0 \exp\left(-\frac{H_{cs}}{3kT}\right) \tag{1}$$

where τ_0 is a constant containing geometric terms and H_{cs} is the activation enthalpy for cross-slip. They represented H_{cs} as $H_{cs} = H^0 - V^*\tau_{010}$ and obtained reasonable values for the enthalpy ($H^0 = 0.3$ eV) and activation volume ($V^* = 6b^3$) from their experimental data.

It is important to note that a critical stress for deformation is inherent in the Takeuchi and Kuramoto model. The model predicts that for stresses less than τ_c dislocation motion will be immediately arrested by the formation of KW locks because the dislocation would not be able to bow out between and break away from the cross-slipped segments. It also predicts that at stresses greater than τ_c the dislocation can easily break away from the cross-slipped segments and move at the free flight velocity.

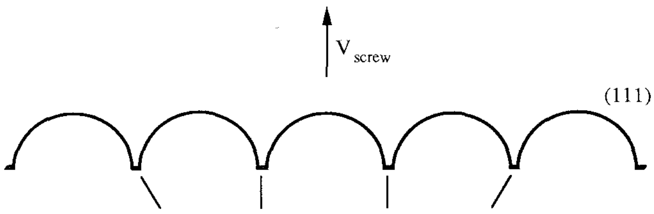

FIG. 1. Schematic of the dislocation structure associated with Takeuchi and Kuramoto's cross-slip pinning model where the formation and subsequent annihilation of pinning points on the (010) plane creates a drag force on the dislocations moving on the (111) plane.

C. The cross-slip pinning model

Experiments by Lall, Chin, and Pope[24] showed Takeuchi and Kuramoto's model to be incomplete. Paidar, Pope, and Vitek,[2] (PPV), modified the Takeuchi and Kuramoto model by deriving an expression for the activation enthalpy for cross-slip which includes (i) the difference in APB energy on both the (111) and (010) planes, (ii) the effect of the applied stress on the dislocation distance between the Shockley partial dislocations, (iii) the Escaig cross-slip effect, as well as (iv) the resolved shear stress on the (010) plane.

Because atomistic calculations have shown that the cores of the superpartial dislocations cannot spread on the (010) plane, the PPV description of the pinning points differs from that originally proposed by Takeuchi and Kuramoto. PPV assume that the leading superpartial dislocation moves only a distance of the order of "**b**" on the (010) plane before redissociating on the (111) or $(1\bar{1}1)$ plane. The double kinked configuration that results from this partial cross-slip process is shown in Fig. 2. PPV suggested that such a cross-slipped segment would act as a pinning point until the remainder of the dislocation bows past it and pulls it back onto the (111) plane. In this way they viewed dislocation motion as being inhibited by the formation and annihilation of an equilibrium number of pinning points. PPV employed the methods of Takeuchi and Kuramoto to relate the frequency of cross-slip to the equilibrium number of pinning points and to the critical stress required for dislocation motion. Like Takeuchi and Kuramoto's model, the PPV model is based on the existence of a critical stress, below which dislocations form KW locks and become sessile and above which dislocations move indefinitely fast.

The PPV model has been compared with the experimental results of Ezz *et al.*[25] and Umakoshi *et al.*[26] and has proven to be very successful in explaining the observed temperature and orientation dependence of the CRSS of Ni_3Al. For this reason, it has been widely accepted as the correct description of cross-slip in Ni_3Al. However, the physical description of dislocations bowing past partially cross-slipped pinning points has recently come under criticism because it is inconsistent with the strain-rate sensitivity that has been measured[27] and because it does not correlate well with the microstructure observed in TEM studies. Vitek and Sodani[8] have modified the PPV model to include the effects of thermally activated unpinning. As a result of this modification, they have been able to use the details of the previous cross-slip pinning models in a way that predicts the observed strain-rate dependence.

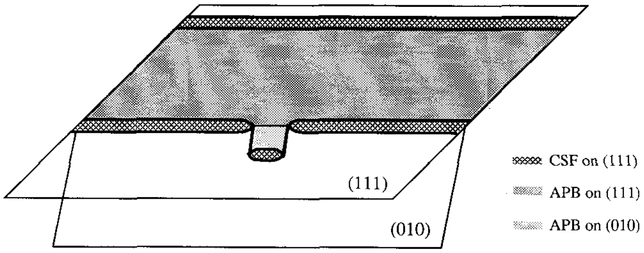

FIG. 2. The cross-slip pinning process as described by Paidar, Pope, and Vitek. Notice that the cross-slipped points are not completely cross-slipped on the (010) plane. In this model they are assumed to be only partially cross-slipped pinning points and as such completely reversible.

The most recent cross-slip pinning model is still unable to explain several key microstructural observations. The vast majority of dislocations that exist after deformation have been observed to be long, relatively straight, screw dislocations. Detailed weak-beam TEM studies[3–5,28–31] have shown that screw segments are completely dissociated on the (010) plane in the form of KW locks, and that these KW locks are connected by short dislocation segments on the (111) plane. This observed microstructure is in direct contrast with the dislocation structure predicted by the cross slip pinning model, i.e., short screw segments pinned on the (010) plane with long segments bowing out between them on the (111) plane. Moreover, the observation of superpartial dislocations widely dissociated and forming APB's on the (010) plane is not predicted by the PPV model, and it can be taken as an indication that the cross-slip process is far more complete than envisioned by the PPV model. It has been suggested that the dislocation structure observed in these TEM studies is actually a relaxed structure and not a true picture of deformation in Ni_3Al, but Korner,[28] Mills *et al.*,[5] and Sun[29] have observed local bowing of KW locks on the (010) plane. This bowing can occur only when the stress is applied and is a strong indication that KW locks are formed during deformation and are not the result of a relaxation process.

D. Superkink models

The TEM observations have led to the proposal of alternate descriptions of dislocation motion. Sun and Hazzledine[3] were the first to suggest that the flow stress in Ni_3Al is controlled by the motion of KW locks by the formation and propagation of double kink segments on the (111) plane. As can be seen in Fig. 3, the lateral motion of these segments will lead to the propagation of the locked screw dislocations. This motion is similar in nature to the double kink process commonly used to explain dislocation motion in hard crystals where a high lattice friction inhibits dislocation motion. Since the lateral motion of these kinks is assumed to be a relatively easy process, flow is thought to be controlled by the rate of formation of double kink segments. Caillard and his co-workers[6,7] have observed the jerky motion of screw dislocations with *in situ* electron microscopy.

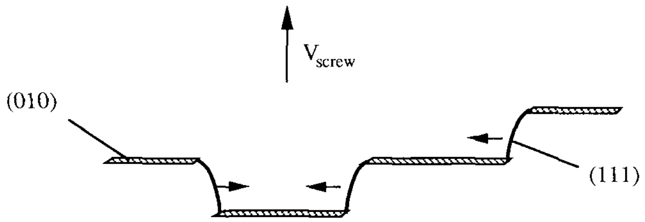

FIG. 3. A schematic of dislocation motion as described by the cross-slip pinning models. KW locks on the (010) planes are believed to be propagated by the lateral motion of dislocation kinks on the (111) plane.

They interpret this jerky motion to be caused by a locking-unlocking mechanism that is similar in nature to the formation and propagation of large superkinks, as was proposed by Sun and Hazzledine.[3] It is not, however, entirely clear that the formation of large double kink segments on the KW locks in Ni_3Al would be energetically feasible, and a complete model for the anomalous yielding behavior based on this process has not been published.

Veyssière[4] also observed that the kinked segments between KW locks lie on the (111) plane. He assumed that this structure was formed during the expansion of dislocation loops on the (111) plane and argued that double kink formation is not necessary for kink motion to control the motion of the dislocation. Veyssière proposed that at higher temperatures localized bowing of KW locks on the (010) plane reduces the superkink mobility and leads to higher yield stresses. But, observations of the anomalous yielding behavior in crystals oriented along the [001] direction suggest that bowing on the (010) plane is not required for the anomalous flow strength. In fact, Dimiduk[30] has reported that the anomalous flow behavior is observed in specimens where the KW locks were not observed to bow out on the (010) plane and has therefore discounted the necessity, although not the effect, of bowing on the cube cross-slip plane.

Mills *et al.*[5] have also suggested that the formation of KW locks during loop expansion and subsequent movement of the kinked segments on the (111) plane leads to the flow stress anomaly of Ni_3Al, but their work differed from the previous models in two important ways. First, they suggested that the average length of the kink segments depends on the frequency of the (010) cross slip event and is therefore dependent on orientation and temperature in the way described by Paidar *et al.*[2] They also suggested that the kink segments were not confined to kink-like motion and proposed that the kink segments would bow out and glide on the (111) plane. The stress required for this bowing process increases as the length of the kink segment decreases, and is also related to the frequency of cross-slip.

Hirsch[9,10] has adopted a view of octahedral glide that is almost identical to that proposed by Mills *et al.*, and he has developed a very detailed model of the dislocation reactions that will occur when superkink segments bow out between and unlock the KW locks. The Hirsch model is currently the most developed and quantitative of the superkink models. It is important to note that, like the cross-slip pinning models, Hirsch's superkink model is dependent on the existence of a critical stress for dislocation motion. In this superkink model, the critical stress arises from the fact that an average kink height is used to calculate the stress required for unlocking and because no provision for strain hardening has been included in the model.

The transient deformation experiments described in this paper have been designed in an attempt to distinguish among the different dislocation models that have been proposed. An effort has been made to identify the basic flow properties of each model and to relate these properties to the results of the transient experiments. The results of these comparisons have led to a more complete description of octahedral glide in Ni_3Al. It is our hope that this description will provide a stronger basis for further model development and improve the understanding of flow in this material.

III. EXPERIMENTAL PROCEDURES

Single crystals (5 mm × 15 mm × 150 mm) of single phase Ni_3Al (Ta), (Ni–24 at.% Al and 1.0 at.% Ta), were seeded with an orientation near the $[\bar{1}23]$ and grown for this study by J. R. Whetstone and D. Frasier of the Alison Gas Turbine Division of The General Motors Corporation. Prior to machining, these crystals were vacuum annealed at 1530 K for 2 h by Dr. J. P. Wittenauer of the Lockheed Missiles and Space Corporation. Parallel piped (4.25 mm × 4.25 mm × 11.75 mm) compression specimens were electrode discharge machined from the crystals. All specimen faces were low-stress-ground, mechanically polished to 25 μm, and electropolished to provide a smooth surface for testing and subsequent slip trace verification. The specimen orientations were measured using a Laue back reflection technique and found to be [−0.95, 2.15, 3].

Constant true strain-rate compression experiments were conducted at 4×10^{-4} (1/s) in the temperature range of 25–500 °C, where the yield strength of $Ni_3Al(Ta)$ increases with increasing temperature. In the temperature change experiments, annealing effects were minimized with the following procedure: the furnace was stabilized at 500 °C, the specimen was introduced and allowed to come to equilibrium (a process which took 2–3 min), immediately after the first deformation the specimen was removed from the furnace and air cooled, the furnace was then stabilized at 100 °C, and the specimen was reinserted and deformed as soon as it came to equilibrium at the new temperature (again 2–3 min). The stress relaxation and stress relaxation/temperature drop experiments were both conducted at 300 °C. Stress relaxation experiments were conducted by fixing the displacement of the crosshead, and the temperature drop experiments were performed while the stress was held constant by computer control.

IV. DEFORMATION TRANSIENT EXPERIMENTS

A. Primary creep

The observation of octahedral glide during primary creep[11] suggests that yielding and primary creep are related processes. As is shown in Table I, the amount of strain accumulated during primary creep is inversely related to the test temperature; it decreases with increasing temperature. This observation combined with the fact that the dislocation substructure associated with primary creep (screw dislocations in the form of KW locks) is the same as for yielding both suggest that the dislocation mechanisms that lead to the anomalous yielding behavior of Ni_3Al are also active during primary creep. It follows that the study of primary creep can be used to provide additional information about the characteristics of octahedral glide in Ni_3Al.

The observation of octahedral glide in creep tests with the applied stress less than 75% of the CRSS suggests that a "critical stress" for dislocation propagation, as originally proposed by Takeuchi and Kuramoto,[1] does not exist. The cross-slip pinning models[2,8] and several of the superkink models[5,9,10] are based on a temperature dependent, characteristic distance which necessarily produces a critical stress behavior. Models such as these which incorporate a critical stress are therefore inconsistent with the observations of primary creep. The initial portion of the creep curves exhibits a creep rate that decreases with increasing strain. This "normal" primary creep response is similar in appearance to the "Class II" type behavior that Sherby and Burke[32] used to describe creep in pure fcc metals and certain solid solution alloys where deformation is controlled by the development of dislocation substructure. This suggests that substructure formation also has an important effect on the flow behavior of Ni_3Al. However, the importance of work hardening and substructure formation are not included in any of the currently accepted models for yielding in Ni_3Al.

TABLE I. Temperature dependence of primary creep in Ni_3Al (Ta) (Applied axial stress = 300 MPa).

Temperature (°C)	ε_p (25 h)
357	0.009
550	0.004

Primary creep does not lead to steady state creep, but is instead followed by the exhaustion of octahedral glide after a period of a few hours. The fact that primary creep leads to exhaustion, and not steady state creep, has been taken as an indication that dislocation motion is arrested by the formation of KW locks. The absence of thermally activated recovery can be interpreted to mean that the formation of KW locks is an irreversible, nonrecoverable process at a fixed stress and temperature.

The discrepancies between the primary creep observations and the proposed models suggest that additional transient experiments are needed to characterize octahedral glide in Ni_3Al. The experiments described in this paper were designed to determine the importance of dislocation substructure and to characterize the nature of dislocation motion.

B. Cottrell–Stokes type experiments

The effects of dislocation mobility and dislocation substructure on the flow properties of a material can be evaluated by the use of a temperature change test first proposed by Cottrell and Stokes.[12] When flow is limited by the dislocation mobility, dislocation glide is a dynamic process and a change in temperature will result in an immediate change in the flow stress. In this way, mobility controlled dislocation glide is a thermally reversible process. However, when flow is substructure dependent, changing the temperature without changing the structure will not lead to an immediate change in the flow stress. In this case, dislocation glide is not a thermally reversible process because the dislocation substructure does not immediately recover when the temperature is changed.

The proposed yielding models for Ni_3Al can be classified by their thermal reversibility. For example, the cross-slip pinning model predicts that an equilibrium number of pinning points exists at a given temperature and that the formation and annihilation of pinning points is completely reversible with temperature. At the other extreme, the source expansion model proposed by Mills *et al.*[5] is very dependent on dislocation structure and assumes that the formation of superkink segments between KW locks is an irreversible process, and that the length of superkinks will not change immediately with temperature.

Cottrell–Stokes type temperature change experiments have been performed to determine the thermal reversibility of octahedral glide in Ni_3Al. Monotonic tests were conducted at 100° and 500 °C and a temperature change test was run such that the specimen was strained 2% at 500 °C, unloaded, and reloaded at 100 °C. It was expected that if dislocation motion were thermally irreversible, then the flow stress would return to its value of 500 °C, but that if dislocation motion were completely reversible, it would drop to the value measured in the 100 °C monotonic test.

The results obtained from the Cottrell–Stokes experiment are given in Fig. 4. As is shown in this figure, the value of the flow stress after the temperature change is equal to the sum of the 100 °C yield stress and the change in stress caused by strain hardening (Θ) between 0.2% and 2.3% plastic strain at 500 °C. This indicates that the "harder" structure introduced at 500 °C influences the subsequent deformation at 100 °C, but that the processes that control deformation are also partially reversible. It is concluded that there is both a reversible and an irreversible contribution to the flow stress.

It is also observed in Fig. 4 that the strain hardening rate anomalously increases with temperature. These constant strain rate tests were conducted on single crystal specimens oriented so that only one {111} slip system was active. Consequently, the observed strain hardening cannot be attributed to interactions with a secondary slip system. It can, however, be explained in terms of the formation of KW locks as was first suggested by Kear.[22] Strain softening was not observed after the temperature change. The absence of strain softening is an indication that the substructure formed at the higher temperature is permanent, and suggests that the assumption that the formation of KW locks leads to strain hardening is correct.

The results of these experiments are in agreement with earlier studies by Davies and Stoloff[33] and more recently by Dimiduk.[30] Davies and Stoloff as well as Dimiduk discounted the effects of strain hardening and have interpreted their results as an indication that yielding in Ni_3Al is a completely reversible process. Ignoring the effects of strain hardening altogether can be questioned. As can be seen in Fig. 5, the anomalous strain hardening rate has an important effect on the measured 0.2% offset yield stress. The effects of strain hardening have been given special attention in the present study; the difference between partially and fully reversible flow will be addressed in greater detail in the discussion of the exhaustion/temperature drop experiments.

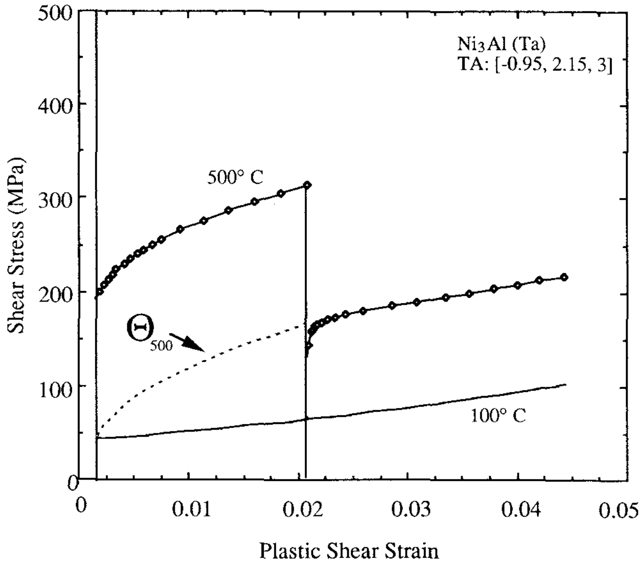

FIG. 4. The results of the temperature change test for a single crystal of Ni_3Al(Ta). The flow stress after the change is observed to be both dependent on strain hardening and partially reversible.

C. Stress relaxation experiments

Materials in which flow is limited by the dislocation mobility are generally acknowledged to be very strain rate sensitive and are usually characterized by a large strain rate exponent (m) and a low stress exponent (n). The drag associated with the formation and annihilation of pinning points and the propagation of dislocations by the lateral motion of superkinks are both processes that limit dislocation mobility. However, the effect of strain rate on the flow stress of Ni_3Al as measured in yielding experiments has been shown to be very small.[27] In order to determine if dislocation mobility plays an important role in determining the flow stress, stress relaxation experiments have been conducted to measure the effects of stress and strain rate on the flow properties of Ni_3Al.

In a stress relaxation experiment the crosshead is stopped during a yielding experiment and the stress acting on the specimen is allowed to relax. The stress is measured as a function of time, and the effect of stress on the strain rate can be determined by comparing

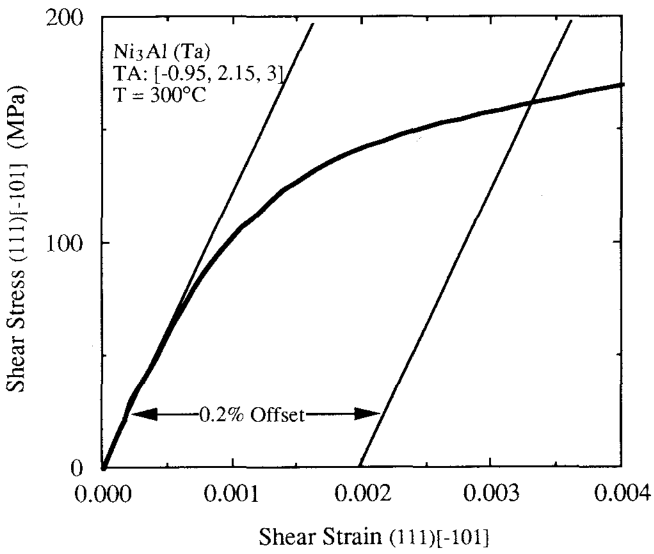

FIG. 5. The stress-strain curve for yielding. The smooth shape of this curve indicates that the flow stress measured at 0.2% offset strain includes the effects of strain hardening. The absence of an abrupt change in this curve suggests that flow is not controlled by lattice friction but is instead dependent on the formation and subsequent recovery of the dislocation substructure.

the change in stress rate with the stress. If the strain rate is modeled as having a power law dependence on stress, then a log-log plot of stress rate versus stress gives the stress exponent (n). If the strain rate is modeled as having an exponential dependence of stress, then a semi-log plot of stress rate versus stress will produce the activation volume (V_{act}). Both n and V_{act} are parameters that describe deformation in the material, and as such they can be used to identify the microstructural mechanisms that control deformation. Low temperature stress exponents (n) and activation volumes (V_{act}) for various materials are given in Table II. From this table, it can be seen that for materials in which dislocation mobility controls deformation (i.e., Class I solid solution alloys and bcc metals) typical values of n and V_{act} are 5 and 100 b^3, respectively. By comparison, materials in which deformation is controlled by the dislocation substructure (i.e., pure fcc metals and Class II alloys) have values for n and V_{act} that are closer to 100 and 2000 b^3, respectively.

The curve given in Fig. 6 is representative of the relaxation curves that were obtained in this study. As can be seen in this figure, the magnitude of the stress relaxation for Ni_3Al is very small (less than 6% of the initial stress). The fact that the stress relaxes quickly, with very little change in stress, suggests that dislocations are immobilized much more completely than predicted by the cross-slip pinning model. This rapid exhaustion is also much faster than the stress relaxation that has been measured for a number of bcc metals and alloys,[34] and it is therefore inconsistent with the idea that the dislocation mobility is controlled by the unrestricted lateral motion of superkinks.

As is shown in Fig. 7, the stress exponent for Ni_3Al was determined to be 200. Comparing this value ($n = 200$) with the different materials listed in Table II indicates that deformation in Ni_3Al is not dislocation mobility controlled but is instead dislocation obstacle controlled. Stoiber *et al.*[35] have conducted similar experiments in order to determine the activation volume for Ni_3Al. Results from their work indicate that the activation volume for Ni_3Al (1% Ta) is about 2000 b^3. This value correlates well with the value of stress exponent (n) measured in the present work.

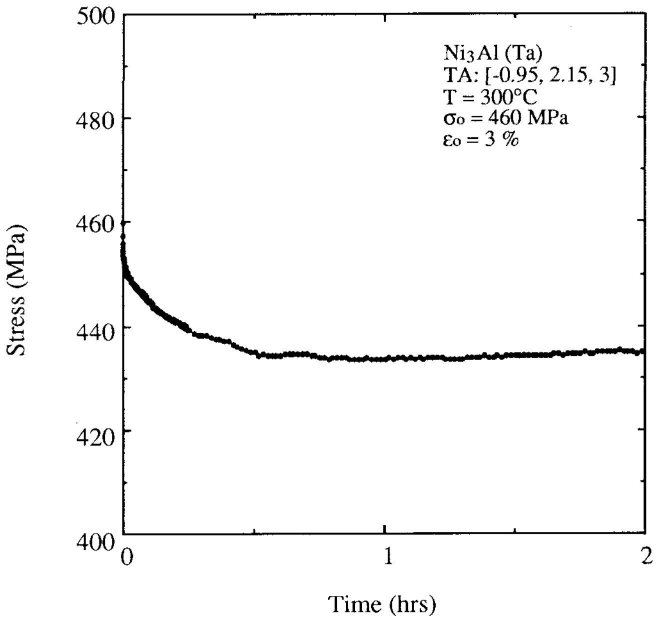

FIG. 6. A stress relaxation experiment for Ni_3Al(Ta) that was conducted after 3% plastic strain. Notice that the relaxation exhausts after only a 6% decrease in stress.

The results of these stress relaxation experiments suggest that the reversible part of the flow strength anomaly is not caused by the drag associated with the formation and annihilation of an equilibrium number of cross-slip pinning points or by the unrestricted lateral motion of superkink dislocation segments. Instead, yielding appears to be related to the rapid motion of a small number of dislocations that are able to move some free flight distance before they cross-slip and become immobilized.

TABLE II. Typical values of stress exponents (n) and activation volumes (V_{act}).

	Material	Ref.	n	V_{act} (b^3)
Dislocation mobility controlled	Nb	(i)	5 – 7	50
	LiF	(i)	5 – 7	140
	W	(i)	5 – 9	5
Dislocation obstacle controlled	Zn	(ii)	50 – 100	. . .
	Cu	(iii)	100	2000
	Ag	(iv)	300	1770

(i) J.C.M. Li, "Kinetics and Dynamics in Dislocation Plasticity", in *Dislocation Dynamics*, edited by A.R. Rosenfield, G.T. Hahn, A.L. Bement, Jr., and R.I. Jaffee (McGraw-Hill, New York, 1968), p. 87.
(ii) K.H. Adams, T. Vreeland, Jr., and D.S. Wood, Mater. Sci. and Eng. **2**, 37 (1967).
(iii) J. Bonneville, B. Escaig, and J.L. Martin, Acta Metall. **36**, 1989 (1988).
(iv) H. Mecking and U.F. Kocks, Acta Metall. **29**, 1865 (1981).

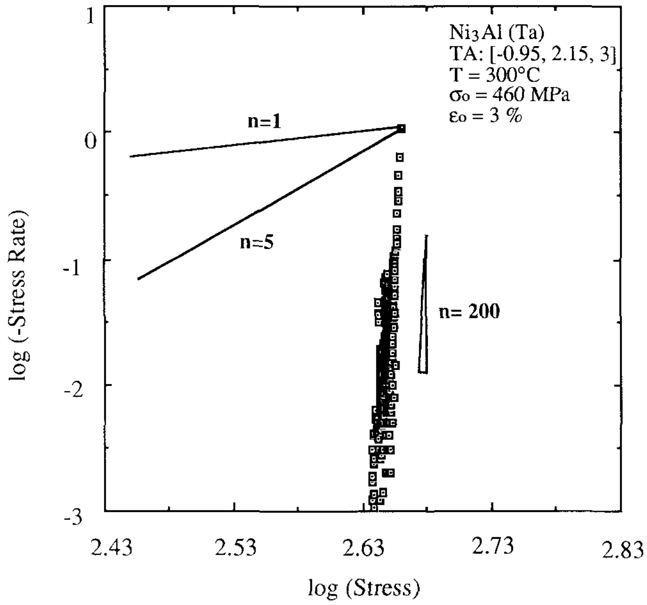

FIG. 7. The low temperature stress exponent of the data in Fig. 9 was determined to be 200. Comparing this value with the data in Table I suggests that flow in Ni_3Al is not dislocation mobility controlled but is instead dislocation obstacle controlled.

D. Exhaustion/temperature drop experiments

Exhaustion/temperature drop tests have been devised to determine if the exhaustions observed during the stress relaxation and creep experiments are due to a complete immobilization of the dislocations or are simply the result of a very low dislocation mobility. If dislocation motion in Ni_3Al is a completely reversible process, then the exhaustion observed in the stress relaxation experiments and during primary creep must be thermally reversible, and a decrease in temperature should result in an immediate initiation of dislocation motion and the observation of plastic straining. However, if dislocation motion is at least partially inhibited by the formation of KW locks, a decrease in temperature will not necessarily lead to an increase in dislocation motion because the formation of KW locks is a thermally irreversible process.

A constant strain rate experiment was conducted at 300 °C with the specimen deformed to 2.0% plastic strain. The crosshead was then stopped and the stress was allowed to relax for 2 h while the stress was monitored. Finally, the test was switched to constant stress control and the temperature was decreased while the strain was monitored. As is shown in Fig. 8, the temperature was reduced by approximately 40 °C before any change in strain was observed. This drop in temperature without a corresponding increase in strain suggests that dislocation motion in Ni_3Al is at least a partially irreversible process.

In a similar experiment, a constant stress of 330 MPa was applied to a creep specimen and the specimen was

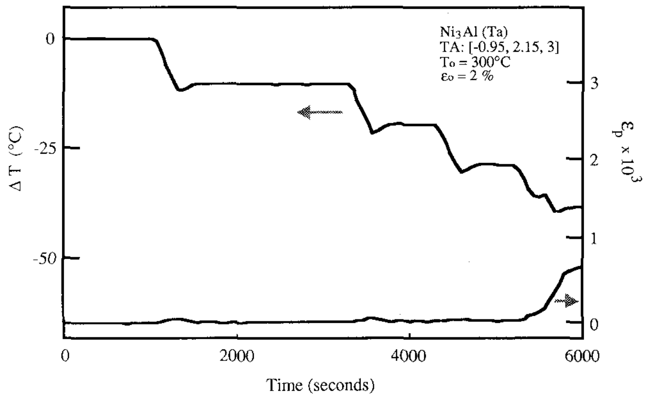

FIG. 8. Deformation exhaustion/temperature drop experiments associated with stress relaxations. The relaxation occurred after 2.3% plastic strain and the observed decrease in temperature without a corresponding increase in plastic strain indicates that flow is only a partially reversible process.

allowed to deform at 300 °C until octahedral glide was exhausted. After this time (24 h) the furnace was turned off and the strain was monitored as the temperature was allowed to drop. A temperature drop of approximately 30 °C occurred before an increase in plastic strain was observed; see Fig. 9. This drop in temperature without a corresponding increase in strain is taken as another indication that octahedral glide is not fully reversible and that the formation of KW locks during primary creep leads to irreversible substructure strengthening.

V. DISCUSSION OF RESULTS

A. The absence of a critical stress for octahedral glide

The relationship between the formation of cross-slipped obstacles and the CRSS originally developed

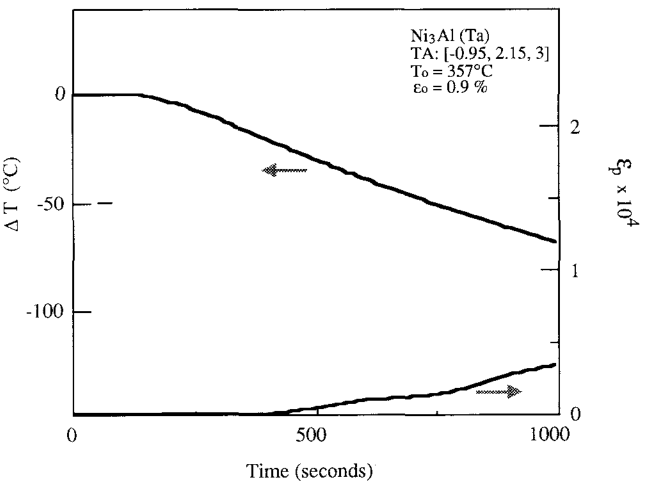

FIG. 9. Deformation exhaustion/temperature drop experiments conducted after the exhaustion of primary creep. The observed decreases in temperature without an increase in plastic strain also suggests that flow is only a partially reversible process.

by Takeuchi and Kuramoto[1] and later employed by Paidar *et al.*[2] and Vitek and Sodani[8] in the cross-slip pinning models assumes that there is a critical stress for dislocation motion. The superkink models of Mills *et al.*[5] and Hirsch[9,10] are also based on the existence of a critical stress for dislocation motion. The observation of octahedral glide during primary creep where the stress is less than τ_c and the absence of a distinct yielding point on the stress-strain curve both suggest that these models, which incorporate a critical stress for flow, do not provide an accurate description of octahedral glide in Ni_3Al.

B. Dislocation mobility versus dislocation obstacle controlled glide

The results of the stress relaxation experiments conducted in this study and in the work of Stoiber *et al.*[35] are in conflict with the assumption that octahedral glide is controlled by either the temporary formation of pinning points or the need for the lateral translation of superkinks. Although it is based on the existence of a critical stress, the description of dislocation motion inherent in the cross-slip pinning model assumes that the vast majority of screw dislocations are moving at some limited average velocity that is inversely dependent on temperature. The propagation of KW locks by the lateral motion of kinked segments is also a mobility controlled process. However, the stress exponent ($n = 200$) measured in this study and the activation volume ($V_{act} = 2000\, b^3$) reported by Stoiber *et al.*[35] both indicate that octahedral glide in Ni_3Al is not dislocation mobility controlled but is, instead, dislocation obstacle controlled.

In metals and alloys where dislocation motion is obstacle controlled (fcc metals and Class II solid solution alloys) deformation is thought to be achieved by the rapid motion of a relatively small number of highly mobile dislocations. In this situation, dislocations are free to glide until they become tangled in the dislocation substructure, at which point the generation of new dislocations is required to continue deformation. In this light, octahedral glide in Ni_3Al is best described as a highly stress sensitive process controlled by the generation and subsequent immobilization of dislocations. A model in which dislocations are generated and able to glide some free flight distance before they cross-slip and become immobilized would be more consistent with the results of the stress relaxation experiments than either the cross-slip pinning model or the superkink model.

C. Substructure formation

The success of the cross-slip pinning model in describing the temperature and orientation dependence of the CRSS suggests that the cross-slip process, as described by the cross-slip pinning model, is somehow integral to the flow behavior. However, a key deficiency in the cross-slip pinning model appears to the lack of provision for work hardening, which also has a strong anomalous temperature dependence. The transient observations that have been made in this study suggest that this deficiency is due to the fact that the cross-slip pinning model assumes that cross-slip segments are only temporary obstacles for dislocation motion.

The results of the Cottrell–Stokes type experiments also suggest that permanent obstacles are formed and that strain hardening plays an important role in determining the flow properties of Ni_3Al. The microstrain measurements of Thorton *et al.*,[36] the observation of a normal primary creep transient, the gradual change in the stress-strain curve, and the results of the Cottrell–Stokes type experiments all indicate that the formation of dislocation substructure inhibits octahedral glide and contributes to the anomalous flow behavior of Ni_3Al.

TEM observations of both creep and yielding specimens indicate that KW locks are the prominent feature of the deformation substructure. The fact that anomalous strain hardening is observed for single slip orientations also suggests that strain hardening is caused by the formation of KW locks. The formation of KW locks should therefore be included in a complete description of the flow strength anomaly in Ni_3Al.

D. Dynamic recovery in Ni_3Al

One of the major objections to the use of KW locks in a description of the flow strength anomaly lies in the fact that the continuous formation of KW locks would lead to the obstruction of octahedral glide. Without adequate recovery processes the formation of KW locks would result in extremely high strain hardening rates and unrealistically high flow stresses. The shape of the stress-strain curve for Ni_3Al, see Fig. 5, indicates that recovery processes are active during the early stages of deformation. The observation of this recovery process at relatively low temperatures indicates that dynamic recovery, not diffusive recovery, is occurring. Dynamic recovery is usually associated with the cross-slip and subsequent annihilation of oppositely signed dislocations, but for Ni_3Al the cross-slipped dislocations are sessile and unable to glide together and annihilate. An alternative description of recovery is needed to explain the flow properties of Ni_3Al. The kink expansion process originally described by Mills *et al.*[5] can account for the recovery of the dislocation substructure associated with the formation of KW locks. They suggested that the kinked segments that lie on the (111) plane between two KW locks can bow out and glide on that plane. In doing so the bowing segments would "unzip" the KW lock and allow the entire dislocation to glide some free flight distance before it cross-slips and becomes locked again. This description of flow is attractive because it allows for the recovery of KW locks and is also consistent with obstacle controlled glide behavior.

E. An alternative description of octahedral glide in Ni_3Al

Since the deformation transients associated with changes in temperature, stress, and strain were found to be inconsistent with the current yielding models for Ni_3Al, the following alternative description of octahedral glide has been developed. It is based on the idea that cross-slip as described by Paidar *et al.*[2] is operative, but that it leads to the creation of relatively permanent KW locks. Furthermore, this description assumes that flow is far more stochastic in nature than is suggested by the cross-slip pinning model. It incorporates the ideas of exhaustion hardening due to the formation of KW locks as originally proposed by Thorton *et al.*,[36] and also provides for dynamical recovery through the conversion of stored dislocations into temporarily mobile ones.

The stress relaxation experiments suggest that an appropriate description of octahedral glide should involve the motion of a relatively small number of highly mobile dislocations that are generated and able to run some free flight distance before they cross-slip and become locked. The expansion of superkink segments on the (111) plane is a process that is consistent with this description. As originally proposed by Mills *et al.*,[5] the continual formation of KW locks and subsequent expansion of the kink segments between these KW locks is a naturally stochastic process. However, in its present form the kink expansion model is inconsistent with the observation of octahedral glide for stresses less than τ_c. This inconsistency is due to the fact that the model assumes that there is a single kink length associated with each deformation temperature. If there were, instead, a more natural distribution of kink lengths, the rate of kink expansion could be related to the statistical formation of very large kinks and the existence of a critical stress for dislocation motion would be removed.

In this picture of octahedral glide, both the relative size of the kinks and the free flight distance that an expanding kink can move are determined by the frequency of cross-slip. If the frequency of cross-slip is assumed to be dependent on temperature and specimen orientation in the manner described by Paidar *et al.*,[2] the effects of temperature and orientation on the flow stress can be described. Moreover, the effect of temperature on the free flight distance and the kink length distribution would account for the recoverable part of the flow stress, and the formation of KW locks during deformation would explain the existence of a dislocation substructure. This substructure would account for the nonrecoverable portion of the Cottrell–Stokes experiment, provide an explanation for the anomalous work hardening behavior, and explain the TEM observations.

VI. CONCLUSIONS

Observations of transient deformation in Ni_3Al have led to the following characterization of octahedral glide in this alloy:

(1) A critical stress for dislocation motion, which is a central feature of the existing models for anomalous yielding in Ni_3Al, is not consistent with the observation of primary creep or the shape of the stress-strain curve.

(2) *The Cottrell–Stokes type experiments indicate* that there is both a reversible and an irreversible contribution to the flow stress.

(3) The results of stress relaxation experiments suggest that the reversible part of the flow stress cannot be described in terms of dislocation mobility but should instead be related to the stochastic motion of a relatively small number of highly mobile dislocations.

(4) Primary creep, the shape of the stress strain curve, and the Cottrell–Stokes experiments all suggest that strain hardening, through the formation of KW locks, plays an important role in determining the flow stress.

(5) A complete description of octahedral glide should include both the formation of KW locks and a provision for the dynamic recovery of these locks.

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REFERENCES

1. S. Takeuchi and E. Kuramoto, Acta Metall. **26**, 207 (1978).
2. V. Paidar, D. P. Pope, and V. Vitek, Acta Metall. **32**, 435 (1984).
3. Y. Q. Sun and P. M. Hazzledine, Philos. Mag. A **58** (4), 603 (1988).
4. P. Veyssière, in *High Temperature Intermetallic Alloys III*, edited by C. T. Liu, A. I. Taub, N. S. Stoloff, and C. C. Koch (Mater. Res. Soc. Symp. Proc. **133**, Pittsburgh, PA, 1989), p. 175.
5. M. J. Mills, N. Baluc, and H. P. Karnthaler, in *High Temperature Intermetallic Alloys III*, edited by C. T. Liu, A. I. Taub, N. S.

Stoloff, and C. C. Koch (Mater. Res. Soc. Symp. Proc. **133**, Pittsburgh, PA, 1989), p. 203.
6. D. Caillard, N. Clement, A. Couret, P. Lours, and A. Coujou, Philos. Mag. Lett. **58**, 263 (1988).
7. D. Caillard, A. Couret, N. Clement, S. Farenc, and G. Molenat, Proc. 9th ICSMA, edited by D. G. Brandon, R. Chaim, and A. Rosen (Freund Publishing, 1991), p. 139.
8. V. Vitek and Y. Sodani, Scripta Metall. **25**, 939 (1991).
9. P. B. Hirsch, Proc. 9th ICSMA, edited by D. G. Brandon, R. Chaim, and A. Rosen (Freund Publishing, 1991), p. 3.
10. P. B. Hirsch, Philos. Mag. A (1991, in press).
11. K. J. Hemker, M. J. Mills, and W. D. Nix, Acta Metall. et Mater. **39**, 1901 (1990).
12. A. H. Cottrell and R. J. Stokes, Proc. Roy. Soc. **A233**, 17 (1955).
13. O. D. Sherby, R. H. Klundt, and A. K. Miller, Metall. Trans. A **8A**, 843 (1977).
14. D. L. Yaney, J. C. Gibeling, and W. D. Nix, Acta Metall. **35**, 1391 (1987).
15. M. Yamaguchi, V. Paidar, D. P. Pope, and V. Vitek, Philos. Mag. **45**, 867 (1982).
16. N. Baluc, H. P. Karnthaler, and M. J. Mills, Int. Phys. Conf., Sec. 2, 463 (1988).
17. V. Paidar, M. Yamaguchi, D. P. Pope, and V. Vitek, Philos. Mag. **45**, 883 (1982).
18. M. A. Crimp and P. M. Hazzledine, in *High Temperature Intermetallic Alloys III,* edited by C. T. Liu, A. I. Taub, N. S. Stoloff, and C. C. Koch (Mater. Res. Soc. Symp. Proc. **133**, Pittsburgh, PA, 1989), p. 131.
19. M. J. Mills, *Proc. Elec. Micro. Soc. of Am.* **47**, 302 (1989).
20. B. H. Kear and H. G. F. Wilsdorf, Trans-AIME **224**, 382 (1962).
21. B. H. Kear, Acta Metall. **12**, 555 (1964).
22. B. H. Kear and Hornbecker, Trans. ASM **59**, 155 (1966).
23. S. M. Copley and B. H. Kear, Trans. AIME **239**, 977 (1967).
24. C. Lall, S. Chin, and D. P. Pope, Metall. Trans. A **10A**, 1323 (1979).
25. S. S. Ezz, D. P. Pope, and V. Paidar, Acta Metall. **30**, 921 (1982).
26. Y. Umakoshi, D. P. Pope, and V. Vitek, Acta Metall. **32**, 449 (1984).
27. G. R. Leverant, W. Gell, and S. W. Hopkins, Mater. Sci. Eng. **8**, 125 (1971).
28. A. Korner and H. P. Karnthaler, Philos. Mag. A **44** (2), 275 (1981).
29. Y. Q. Sun, Ph.D. Thesis, Oxford University (1990).
30. D. Dimiduk, Ph.D. Thesis, Carnegie Mellon University (1989).
31. N. Baluc, Ph.D. Thèse No. 886, École Polytechnique Fédérale de Lausanne (1990).
32. O. D. Sherby and P. M. Burke, Prog. Mater. Sci. **13** (7), 384 (1967).
33. R. G. Davies and N. S. Stoloff, Trans. AIME **233**, 714 (1965).
34. I. Gupta and J. C. M. Li, Metall. Trans. **1**, 2323 (1970).
35. J. Stoiber, J. Bonneville, and J. L. Martin, Proc. 8th ICSMA, edited by P. O. Kettunen, T. K. Lepistö, and M. E. Lehtonen, (Pergamon Press, Oxford, 1988).
36. D. H. Thornton, R. G. Davies, and T. L. Johnston, Metall. Trans. **1**, 207 (1970).